

THE B PARAMETER IN THE CREEP
OF CADMIUM - MAGNESIUM ALLOYS

JAMES C. BELLAH

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OF CADMIUM - MAGNESIUM ALLOYS

JAMES C. BELLAH



THE B PARAMETER IN THE CREEP
OF CADMIUM - MAGNESIUM ALLOYS

by

James Crawford Bellah
Lieutenant, United States Navy

Submitted in partial fulfillment
of the requirements
for the degree of
MASTER OF SCIENCE
IN
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Thesis
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IN
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PREFACE

Some thirty or forty years ago, design engineers began to increase temperatures and pressures for the operation of power generating and similar equipment. They discovered that marked economies resulted in this evolution. This trend toward higher and higher temperatures and pressures, however, has produced many pitfalls in the path of engineering. Engineers have found that a phenomenon exists in which properly designed and operated machine components, instead of deforming elastically, deform plastically under surprisingly low loads at the elevated temperatures. This phenomenon has been termed "creep" and may be defined as the time and temperature dependent part of the deformation which accompanies the application of stress to a material.

Many creep equations have been presented which have from a practical viewpoint the ultimate goal of relating short time creep tests with long time operational conditions. Unfortunately, by and large, these expressions are found to hold only over limited ranges of the test variables. Investigations at the University of California, Berkeley⁽¹⁴⁾, within the past decade have resulted in the development of a high temperature creep equation which was shown to hold over the entire creep curve



under conditions of constant stress for a number of alloys.*
This expression gives the creep rate as:

$$\dot{\epsilon} = Se^{-\Delta H/RT} e^{\beta\sigma} \quad (1)$$

This investigation is concerned with the study of the "B" parameter in the above equation for a cadmium-magnesium series of solid solutions following the techniques developed at the University of California for aluminum and its solid solution alloys. The experimental work was carried out at the U. S. Naval Postgraduate School from October 1954 to May 1955.

Opportunity is hereby taken to acknowledge the guidance and timely advice given by Professor Alfred Goldberg during the course of this investigation. To Professor Frederick Coonan and Professor John R. Clark, the author wishes to express his thanks for their interest and advice in the many questions that arose in the project. The author also wishes to express his gratitude to the following people whose contributions made this thesis possible: Mr. Arthur Allten for his skillful machining and technical advice in the design and production of the extensometer used; Mr. William Vienna for his aid in recording data and for technical aid; A. H. Medbury, Commander, U. S. Navy and B. R. Naczkowski, Lieutenant Commander, U. S. Navy, for their cheerful cooperation in conducting the experi-

*Non-precipating type

**In the literature, the Greek letter β is frequently used for this parameter which is used here only in the equations.



mental work, and to his wife Mrs. Phyllis Bellah for her cheerful acceptance of the hardships encountered during the preparation of the thesis.



ABSTRACT

Experimental investigations have shown that the high temperature creep rate for many alloys may be stated as:

$$\dot{\epsilon} = S e^{-\Delta H/RT} e^{\beta \sigma}$$

In this work, "B" was found to be independent of the creep structure for a given material and to decrease with an increase of the percent of alloying of Mg in Cd. solid solution.



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TABLE OF SYMBOLS AND ABBREVIATIONS

(Listed in the order of their use in the Text)

$\dot{\epsilon}$	- Strain Rate, in/inch-hour
A	- Constant
e	- Base of the natural logarithm
ΔH	- Experimental constant commonly called an activation energy, CAL/MOL
R	- The Gas Constant , CAL/MOL / °K
T	- The Absolute Temperature in °K
B'	- The area of the flow unit times the displacement necessary for activation divided by the Boltzmann constant
σ	- True stress (load divided by instantaneous area), psi
σ_0	- Yield stress, psi
G	- Shear Modulus, psi
k	- Boltzmann constant
V	- Volume in which a stress fluctuation occurs ,
λ	- Constant depending on the length of the activated dislocation loop
B	- The slope of the logarithm of the strain rate vs the stress curve divided by 2.303



CHAPTER I

INTRODUCTION

In selecting the proper material for high temperature use, the machine designer finds that the mechanical properties of the materials as determined at room temperature are of little importance. The designer must select the material which provides the specific properties required by the machine element. The specification for the alloy may require varying degrees of corrosion and oxidation resistance, strength and hardness, impact resistance, dimensional stability, and other properties peculiar to the end product. Most of the above properties may be evaluated by short time tests and where applicable, extrapolation for long periods of time. As yet, however, there is no generally accepted short-time-creep test evaluation.

The high operating temperatures and speeds of such equipment as gas turbines and steam turbines in recent years have further intensified the importance of short-time evaluation creep tests. It is entirely possible that the "perfect material", as far as other mechanical properties are concerned, may fail after only a short operating life due to creep. There is, therefore, an urgent need for the investigation of the factors affecting the creep of structural materials, from which more definite assumptions and empirical laws controlling the process may be deduced.



With the increase of interest by scientists and engineers in creep and its effects, Schwöpe and Jackson⁽¹⁰⁾ made a survey of the field of creep and have summarized the existing theories relating to creep phenomena and the extent of the current knowledge on the subject. As creep is a rate process which is temperature dependent, much of the work has been done by considering the process as that of activation of flow units analagous to the implications in Eyring's⁽⁵⁾ "Chemical Reaction-Rate Theory". Kauzmann⁽⁶⁾ extended Eyring's analysis to the application of creep in metals arriving at the relation:

$$\dot{\epsilon} = A e^{\frac{-\Delta H}{RT}} \sinh \frac{\beta' \sigma}{T} \quad (2)$$

Deformation or flow in solids in general, is caused by the tendency of the atoms or molecules in an elastically stressed body to adjust in such a way that if the deformation remains constant a release of stress would occur. In metals, deformation causes a mutual displacement of atoms and the system gains potential energy, resulting in internal stresses being set up to balance the external forces. Whenever a portion of the internal stress is continually relaxed, the equilibrium with the external force is disturbed, resulting in flow. Becker⁽²⁾ proposed the theory that the local stress fluctuations due to thermal motion must overcome the local yield stress. He presented the following formula which relates the creep rate, stress, and temperature:

$$\dot{\epsilon} = A e^{-\left[\frac{V(\sigma_0 - \sigma)^2}{2GKT} \right]} \quad (3)$$

In order to meet certain objections against Becker's simple theory, Orowan⁽⁸⁾ replaced σ by $q\sigma$ where q represents the stress magnification at imperfections inside the material. Moreover, in order to introduce the effect of strain hardening, a function $\dot{\epsilon} = f(\sigma, \epsilon, T)$ was assumed in which σ was assumed to be a simple function of the strain $\sigma = \phi(\epsilon)$. The revised formula became:

$$\dot{\epsilon} = A e^{-\frac{V(\sigma_0 - \phi(\epsilon) - q\sigma)}{2GKT}} \quad (4)$$

Subsequent investigations by Andrade⁽¹⁾ attempted to use the modified equation to correlate his experimental results on zinc and cadmium, but were unsuccessful for any reasonable form of $\phi(\epsilon)$. More recently, Corn, Goldberg and Tietz⁽⁴⁾ and additional work by Orowan⁽⁹⁾ have served to disprove the general validity of the above formulation.

The second important mechanism for creep is the movement of dislocations as suggested by Orowan⁽⁹⁾ and more completely described by Taylor⁽¹⁵⁾. This relatively new theory has gained wide popularity and is described by many authors. A dislocation is the result of an imperfection in the lattice network which causes a misfit of atoms. Seitz and Read⁽¹¹⁾ and Bragg⁽³⁾ have thoroughly described the origin and basic prop-



erties of dislocations. In this model, creep arises from thermal activation of dislocations over energy barriers in the slip plane. Under these conditions the Kauzmann⁽⁶⁾ type of relationship is:

$$\dot{\epsilon} = A e^{-\frac{\Delta H}{RT}} \sinh \frac{\lambda \sigma}{RT} \quad (5)$$

More recently a third theory was introduced which considers that creep arises from stress relief at a Frank-Read source as a result of dislocation-climb to a new slip plane. Mott⁽⁷⁾ has suggested that the following relationship holds under these conditions, where ΔH is specifically that for self-diffusion

$$\dot{\epsilon} = A e^{-\frac{(\Delta H - \lambda \sigma)}{RT}} \quad \text{where } \frac{\lambda \sigma}{RT} \gg 1 \quad (6)$$

Sherby, Orr and Dorn⁽¹³⁾ in their extensive investigations on the creep of high purity aluminum and some of its solid solution alloys found that Eq. (2) above was not valid when examined over a wide range of variables of stress, temperature and strain. These writers state that there exists two primary reasons for the failure of this equation:

1. In most testing techniques, the investigators have been unable to isolate the effects of stress, temperature, and the other parameters B , ΔH , and A .

2. It has been proved that the stress term for high temperature enters the analysis as σ and not as σ/T

At this time it can be definitely stated that there is no



general relation governing creep under all conditions which is unequivocally accepted by the scientific world. Success can come only by complete isolation of all the parameters affecting the creep of metals under wide ranges of temperature and stresses.

In recent years, Dorn, Sherby, and their associates have contributed a great deal to the knowledge of creep. Their study⁽¹⁴⁾ of the effect of stress on the creep rate, conducted under conditions of constant structure, have indicated that the following general relation holds:

$$\dot{\epsilon} = S_0 e^{\frac{-\Delta H}{RT}} e^{\beta \sigma} \quad (7)$$

These writers have investigated the various parameters appearing in this relation. Their technique involved in the investigation of the B parameter with which this thesis is concerned, consisted essentially of reducing the other variables to constants and known values.

Their procedure was as follows:⁽¹⁴⁾

A specimen was precrept at a given engineering stress and temperature to a selected engineering strain at which time the true stress as determined from the instantaneous area was reduced to some lower value. The instantaneous true creep rate following reduction of the true stress was then determined.



Second, third, etc., tests were conducted under identical conditions except the stress was reduced to yet lower values on reaching the precreep strain, yielding successively lower instantaneous creep rates. Inasmuch as the precreep conditions were identical in each series of tests, the instantaneous structures present immediately after reducing the stress were presumed to be identical.

Using this technique, the logarithm of the strain rate was plotted against the stress. Straight lines were obtained, the slope of which gave the value of the parameter B. The following conclusions concerning the B parameter were arrived at for aluminum alloys.

1. That B is independent of the precreep strain and therefore of the structure developed during creep.

2. That B is independent of various structures that are developed over a range of precreep conditions.

3. That B also appears to be independent of the test temperatures.

4. That B decreases almost linearly with the atomic percentage of the solute element. It appears prudent to note, however, that the alloying with aluminum did not exceed 2% in this series of tests.

The purpose of the present study was an attempt to verify the conclusions listed above on other alloys, and in addition to extend the investigation over a wider range of alloying percentages. The same procedure was followed, with the exception



that engineering strains were used throughout the study, in lieu of true strains. This should not involve any complications as comparisons were made by extrapolation to a structure for a given strain obtained under identical precreep conditions. The error due to using engineering strain rates instead of true values is considered negligible in the low strain regions employed in this investigation. For example, at a strain of 2% the difference between the true and engineering strains is less than 1%. Furthermore, the difference in the strain rates would be still less (the changes in stress were made at a strain of 1.75%).

As will be discussed later, it was found that facts 1, 2, and 4 stated above regarding the B parameter also held for the cadmium and cadmium-magnesium alloys examined by this writer. The effect of temperature on the B parameter was not investigated.



CHAPTER II

Equipment

At the time the subject for this thesis was conceived, there were available two commercial Baldwin Lima-Hamilton Creep Testing Machines in the Metallurgy Department of the U.S. Naval Postgraduate School. These machines were essentially heavy load machines, normally used in long-time creep testing of high strength material. The machines were equipped with a mechanical extensometer which measured the extension over the total specimen. In view of the data required and the material and specimens selected, it was considered essential that a counterbalance system be installed on the lever arm and that a more accurate extensometer be developed. Fig. 1 shows the general arrangement of the testing machine including the required counterbalance.

A standard nichrome-wound tube furnace manufactured by the Marshall Products Co. was used. Uniformity of temperature control along the length of the furnace was obtained by shunts, consisting of ohmite resistors, tapped into the winding at one-third and two-thirds of the way down the furnace walls. Temperature measurements were obtained by the use of chromel-alumel thermocouples. The power input was regulated by means of a variac in series with the furnace winding. A Leeds and Northrup Micromax controller served to keep the temperature constant. A shunted resistance operated to decrease the amplitude of the



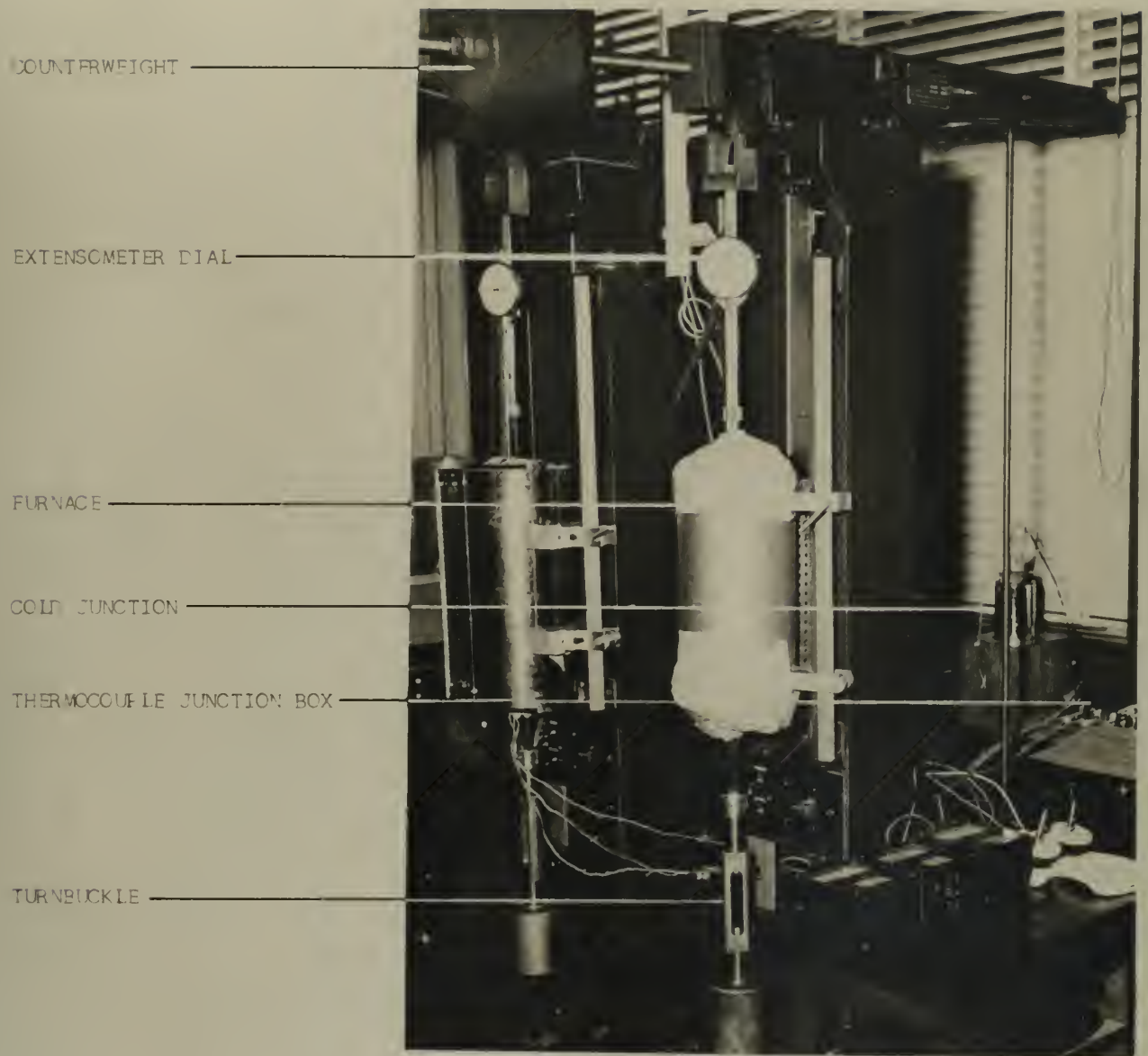


FIG. 1 GENERAL ARRANGEMENT OF TESTING MACHINE



temperature cycling to $\pm 1^{\circ}\text{C}$ with a variation of less than 1°C along the gage length. The temperature readings along the gage length were obtained by three thermocouples which were bound to the specimen with glass string. (See Fig. 2).

The extensometer, as shown in Fig. 3, was designed in cooperation with A. H. Medbury, Commander, U. S. Navy. The extensometer consisted of a concentric sleeve and rod mechanism actuating the movement of a commercial dial indicator. The limit of movement of the dial indicator was one inch. The dial indicator was calibrated in 0.001 in. per division. The sensitivity of the extensometer was checked by loading an aluminum specimen on the 120,000 psi Riehle Testing Machine in the Mechanical Engineering Department. A Baldwin SR-4 load cell was mounted in tandem with the extensometer to insure accurate load readings (See Fig. 4). The sensitivity of the extensometer by interpolation, was determined to be 0.0002 inch when compared to the two SR-4 strain gages mounted on opposite sides of the calibration specimen. The SR-4 gages indicated that little or no bending was exerted on the specimen by the extensometer system.

The attachment of the extensometer to the specimen was made by spring loaded-hardened points which penetrated the surface 0.015 in. at each end of the gage length. The two-inch gage length was accurately determined by the use of a gage block when installing each specimen. The penetration of the gage points did not produce any significant stress concentration, as none of the



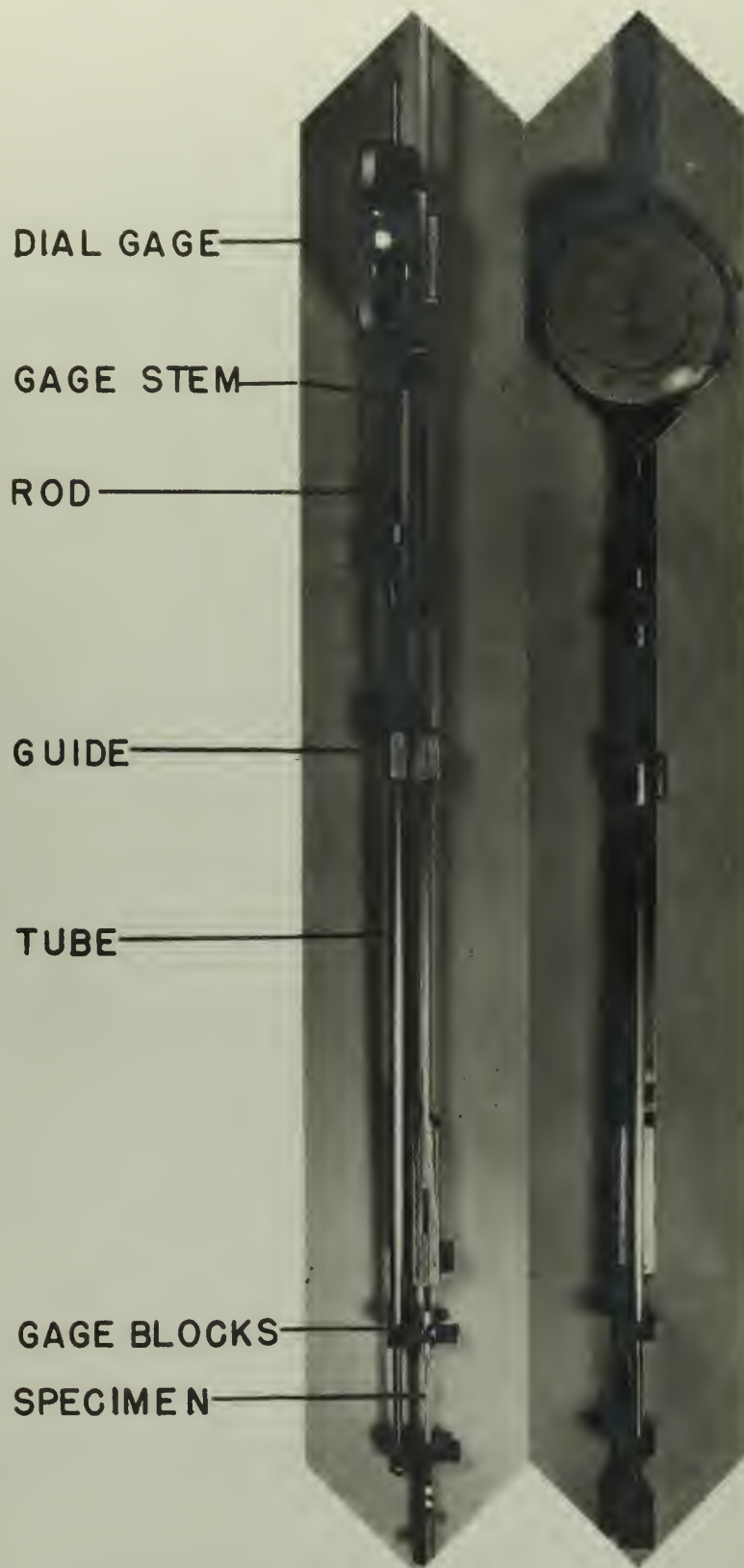


FIG.3 THE EXTENSOMETER



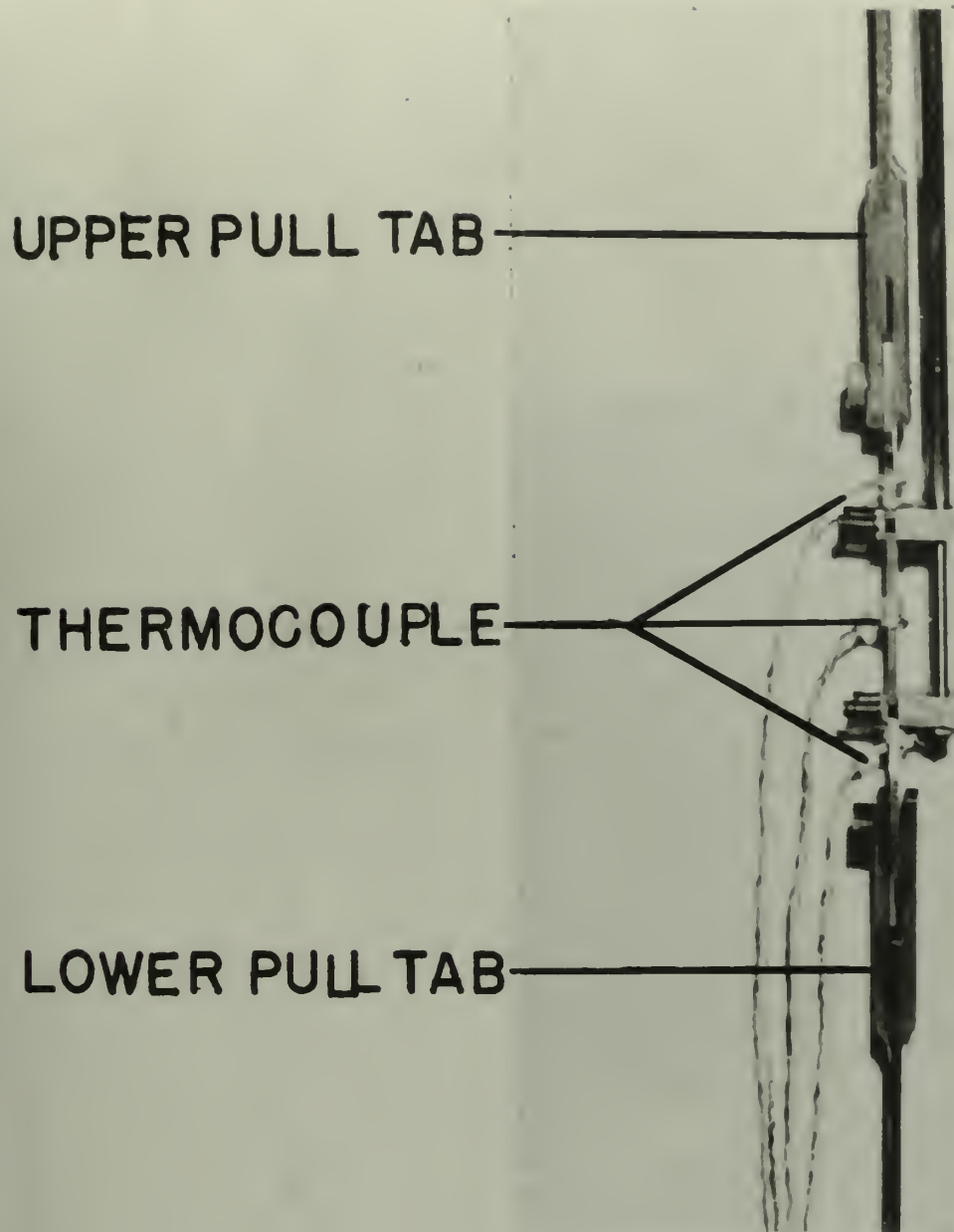


FIG. 2 THERMOCOUPLE-SPECIMEN
ARRANGEMENT



PIDDFLE TESTING MACHINE

LOAD CELL

EXTENSOMETER GUAL

SPECIMEN

STRAIN INITIATION, INFE M

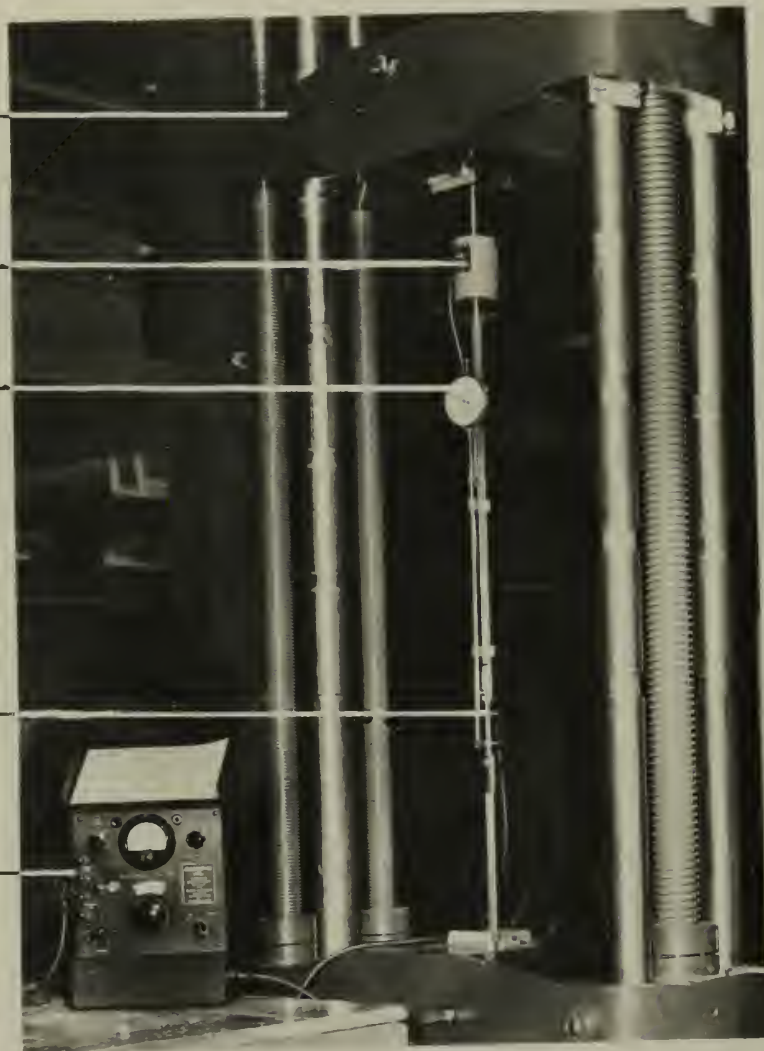


FIG. 4 CALIBRATION OF EXTENSOMETER



specimens which were carried to rupture failed in this area.

The 15 to 1 nominal mechanical advantage of the lever arm of the testing machine was checked on each machine. The Baldwin SR-4 load cell was again used for this purpose. It was determined that the actual mechanical advantage was 14.75 to 1 for each machine, over the entire range for which the machine was used.



CHAPTER III

MATERIAL

High purity cadmium and a series of solid solution alloys of magnesium in cadmium were used for this investigation. The composition of each alloy is given in Table I. The material was procured from the Dow Chemical Co. in the form of 0.1" x 0.75" extrusions. The cadmium had been procured by the Dow Chemical Co. from the Belmont Smelting Co. The Belmont Co. supplied only the weight percent of the Cd in the alloy and did not determine the actual concentration of impurities. However, microstructural studies of the alloys revealed that the small percentage of impurities were in solid solution.

TABLE I
ALLOY DESIGNATION AND CHEMICAL ANALYSIS

CODE NUMBER	NOMINAL ATOMIC PERCENT #	
	CADMUM	MAGNESIUM
70	99.9%	0
71	91.45%	8.45%
72	74.75%	25.15%
73	47.75%	52.15%
69*	0	99.87%

*Impurities in No. 69

Al = .012%; C_a .01%; C_u .001% Fe .001% M_n .001%;
N_i .001%; P_b = .005% S_i = .022% S_n .01%; Z_n = .052%

An approximate composition



A preliminary micrographic investigation of the material revealed that the grain size decreased directly with increasing magnesium content. In addition it was noted that cadmium recrystallizes at room temperature. It was thus decided that the grain size of cadmium would be considered as the standard. It was, therefore, necessary to anneal the remaining alloys in accordance with the schedule in Table II to arrive at the approximate standard grain size. The samples were polished and then etched with a 10% nital solution. The grain count was made by averaging the counts of a number of traverses taken across a section.

TABLE II
GRAIN SIZE OF ALLOYS

Code No.	Metal or Alloy	Annealing Temperature	Time	Approximate Grains/inch
70	Cd	As received	-	80
71	Cd-Mg	175°C	1 hour	100
72	Cd-Mg	200°C	1 hour	100
73	Cd-Mg	300°C	1 hour	100

Each specimen was coated with silicon oil (Dow-Corning 500) to reduce surface oxidation during the test. Excessive

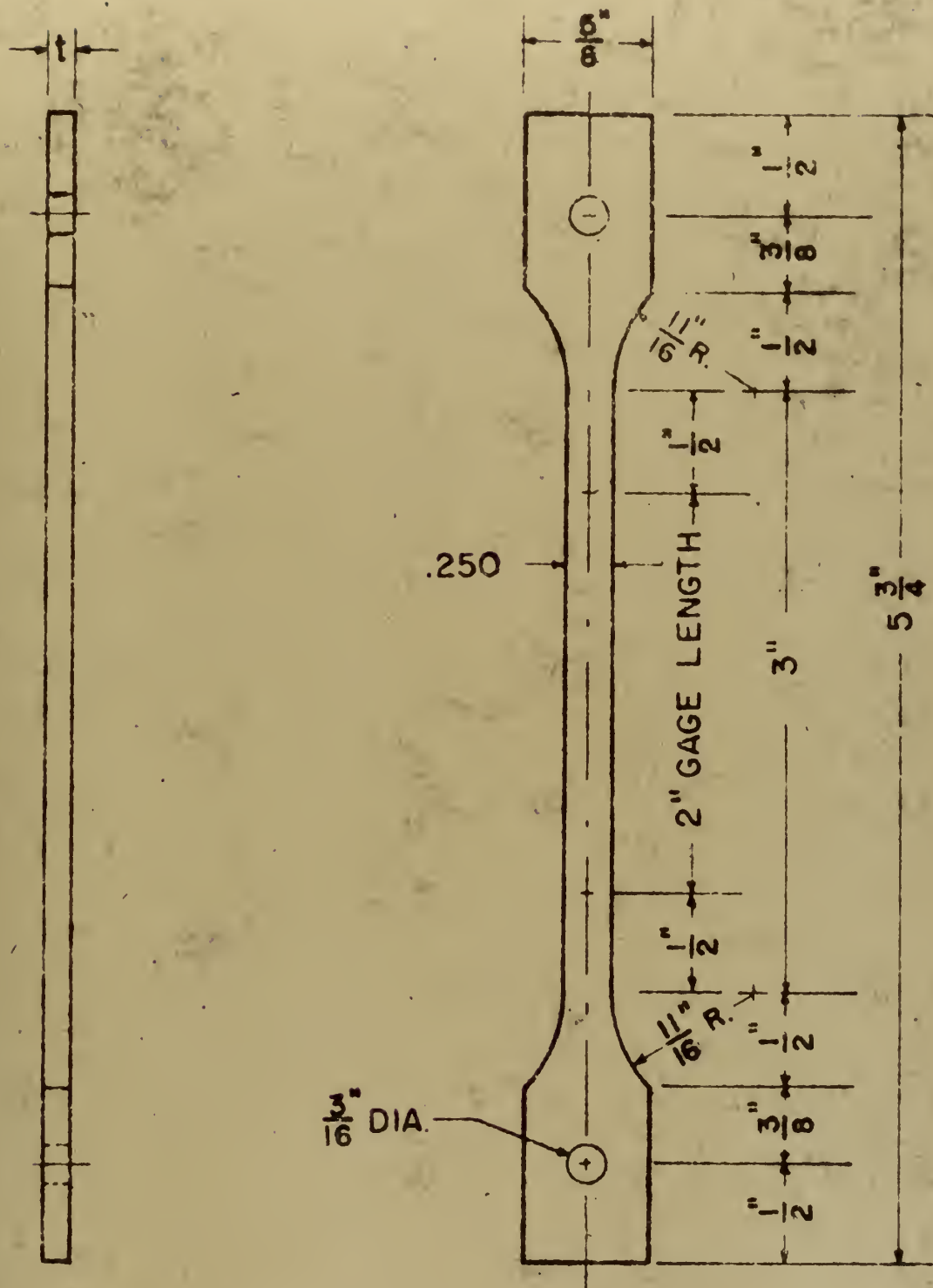


oxidation was noted in alloy 72. However, the results of the tests of this alloy fell in line with the other alloys and are therefore included in the thesis results.

All creep specimens were machined so that their tensile axes were in the direction of extrusion. As specified in Fig. 5, the gage section was 2.0 inches long, 0.25 inches wide, and with a nominal thickness of 0.1 inches. The specimens were prepared by the machine shop of the U. S. Naval Postgraduate School.



FIGURE 5 - Specimen Specifications





CHAPTER IV

PROCEDURE

The lateral dimensions of each specimen were measured with a micrometer to 0.0001 inch in order to accurately determine the cross-sectional area. The weights were measured on a scale to an accuracy of 0.01 pound. It was considered that this permitted the computation of the stress to within 20 psi.

To prevent bending of the specimen and to insure proper seating of the gage pins and exactness of the gage length, the specimen was installed in the extensometer with the aid of a jig which secured the extensometer. The extensometer and specimen were then placed in the testing machine and the thermocouples attached; the furnace was then slid into the proper position. To reduce heat loss, the furnace was thoroughly sealed by the use of aluminum foil, felt, and masking tape. Approximately one hour was required to insure equalization of temperatures in the furnace and specimen. The load was then applied in an interval of about five seconds in a manner to prevent any impact loading.

As described in the introduction, each specimen was then precrept at a given stress (constant load) and constant temperature to a selected engineering strain, at which time the stress was reduced to some lower value and the new creep rate was determined. All of the creep curves exhibited straight lines over



the initial intervals of creep following a reduction of the stress. Consequently, fairly reliable initial creep rates were obtained for the various reduced stresses.



CHAPTER V

EXPERIMENTAL RESULTS

The first group of tests were conducted on specimens of pure cadmium to determine the effect of structure on the B parameter. Fig. 6 presents the creep curves for one series of tests on pure cadmium. Each specimen of this series was precrept to an engineering strain of 1.75% at an initial stress of 2000 psi. When the selected value of precreep was reached, the load was reduced to give a new engineering stress of 1800 psi etc. as shown in Fig. 6. The instantaneous creep rates were measured from this set of curves. The reproducibility for the creep curves for these specimens was considered above the average. The creep rates for the remaining tests are shown in Table III.

Since the precreep conditions of loading and temperature were the same for each specimen in this group, it was assumed that all specimen were in an identical structural state at the instant of reduction of stress. This conclusion was confirmed for similar tests on aluminum alloys by x-ray and metallographic investigation conducted by Sherby and Dorn⁽¹²⁾. The second group of specimens of pure cadmium were precrept to an engineering strain of 5% at an initial stress of 3000 psi.



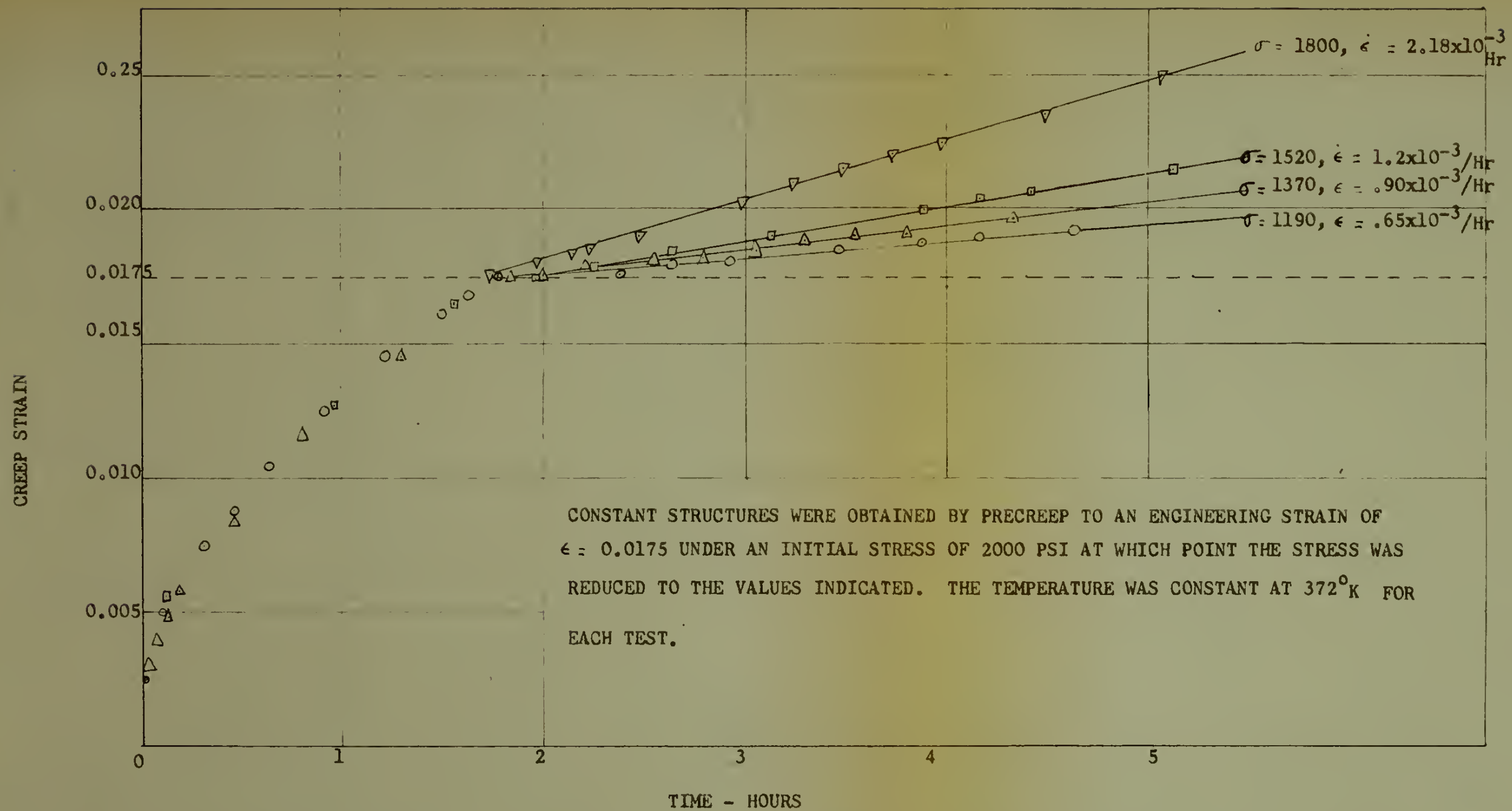


FIG. 6 TYPICAL CREEP CURVES FOR HIGH PURITY CADMIUM SHOWING CREEP RATE DEPENDENCE ON THE STRESS UNDER CONSTANT STRUCTURE CONDITIONS



TABLE III
CREEP RATE DATA

Alloy No.	Initial Stress P.S.I. (Engrg)	Precrept to	Reduced Stress P.S.I. (True)	Creep Rate $\times 10^3$ per Hour	Temper- ature (°K)
70	3,000	.05	1230	1.5	372°
			1535	3.40	
			1835	5.15	
			2035	9.0	
			2465	12.5	
71	3,000	.0175	1420	0.87	372°
			1680	1.23	
			1850	2.00	
			2035	2.06	
			2290	3.86	
72	3,000	.0175	2460	4.75	372°
			1630	1.20	
			1850	1.45	
			2160	2.88	
			2340	3.40	
73	10,000	.0175	2445	4.62	372°
			2645	5.9	
			2900	10.1	
			6040	0.40	
			6950	0.65	
			7570	0.85	372°
			8230	1.40	
			8270	1.50	
			9560	3.45	



Fig. 7 presents a plot of the logarithm of the strain rates versus the stresses for these two series of tests. As shown by the parallel nature of the two lines, the slopes of the curves are independent of the precreep strain. The B parameter therefore, would appear to be independent of the precreep strain and of the structure developed during the creep of pure cadmium. Insufficient time was available to examine this further; however, due to the extensive work done on aluminum and its solid solutions in this connection, it was felt justified to accept the same conclusions here, at least to the extent where the present results are affected.

In order to determine the effect of alloying on the B parameter of cadmium, similar tests were conducted on alloys 71, 72, and 73*. Each of these specimens were precrept to an engineering strain of 1.75% at the initial stresses shown in the applicable figures, Figs. 8 and 9.

The plots of the logarithm of the strain rate versus the stress for the four alloys are shown in Figs. 8 and 9. For convenience and to take advantage of a different scale, the data for alloy 73 was first plotted in Fig. 9 and the slope was then rescaled for plotting in Fig. 8. This permitted a comparison of the effect of the four compositions on the B parameter.

It was noted that the slope of the curve for each alloy decreased with increased percentages of magnesium. This experimental fact was in complete concurrence with the conclusions of

*Described in Table I

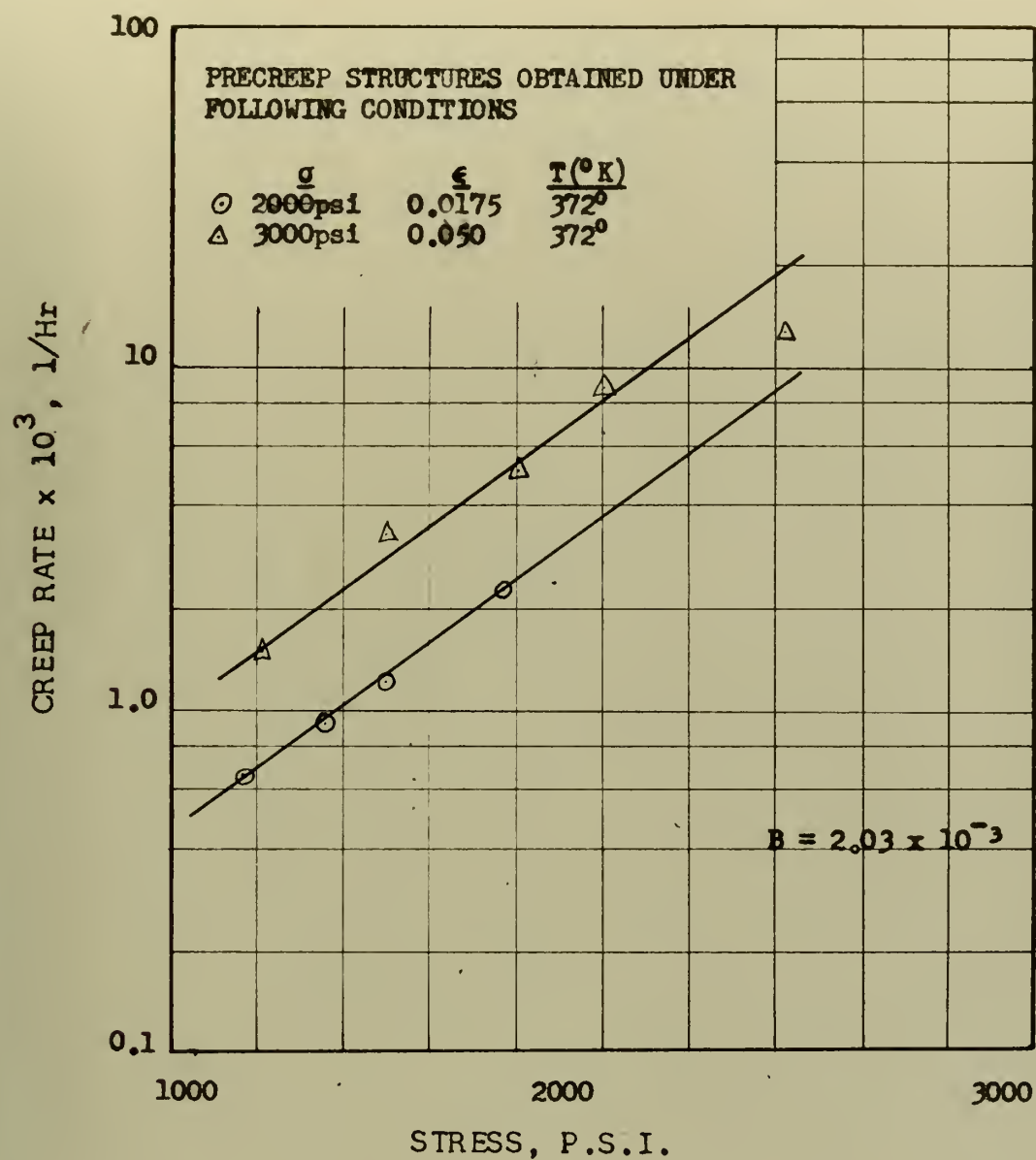


Fig. 7 EFFECT OF STRESS ON THE CREEP RATE AT CONSTANT STRUCTURE FOR HIGH PURITY CADMIUM



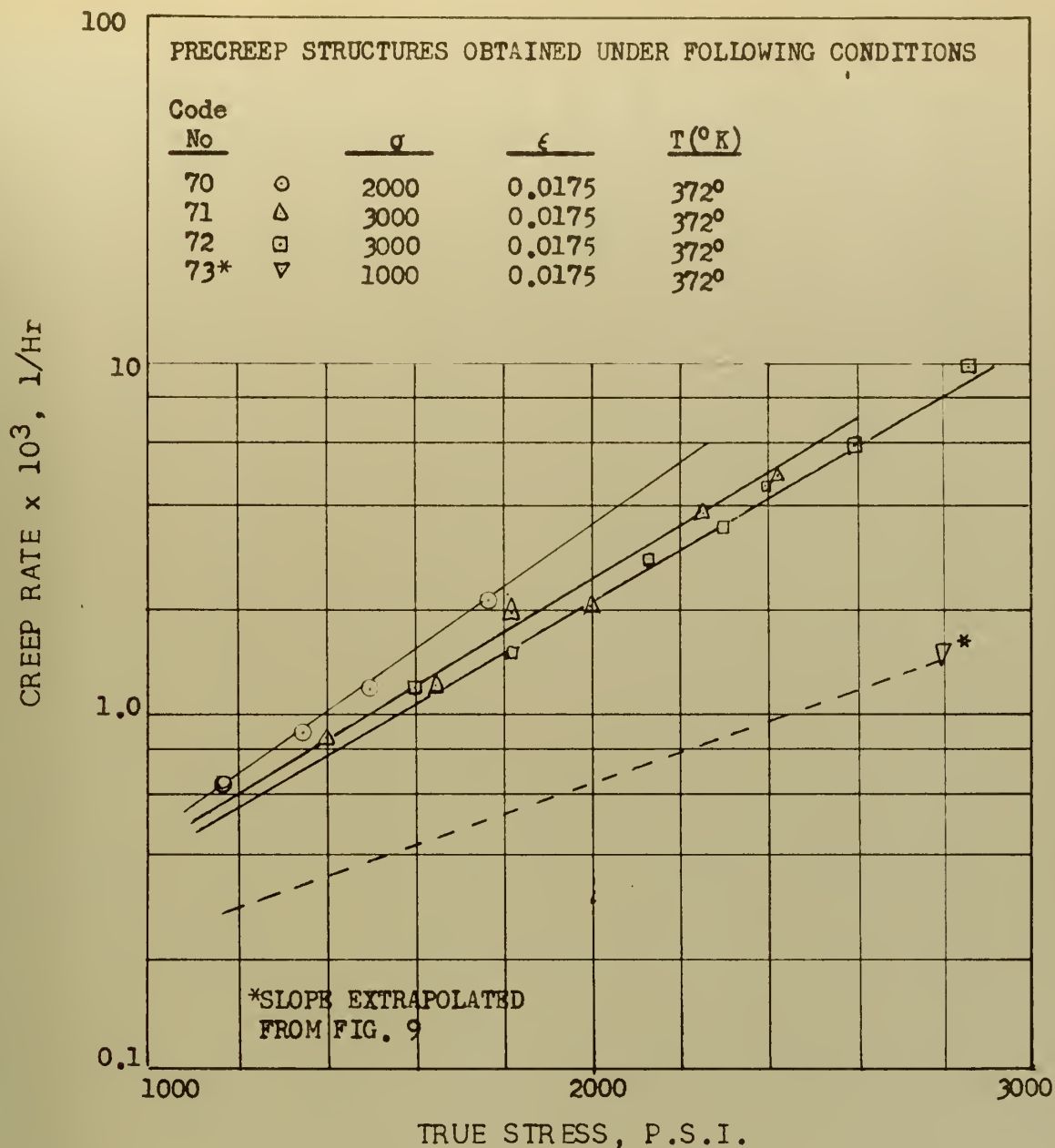


Fig. 8 EFFECT OF STRESS ON THE CREEP RATE
AT CONSTANT STRUCTURE FOR CADMIUM-MAGNESIUM
ALLOYS



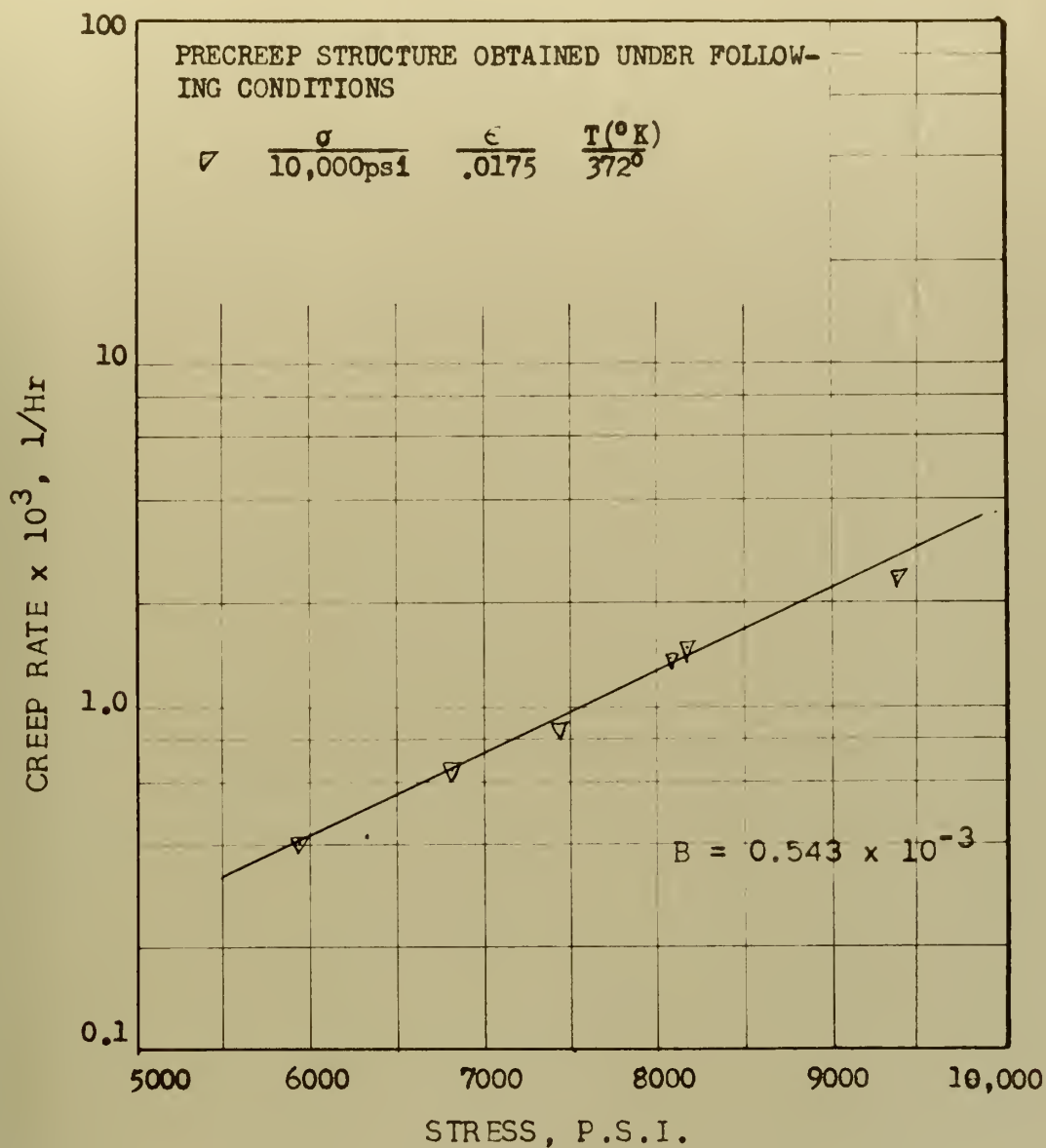


Fig. 9 EFFECT OF STRESS ON THE CREEP RATE FOR CONSTANT STRUCTURE FOR ALLOY



Sherby and Dorn for aluminum alloys. Fig. 10 presents the effect of varying the alloy content on the B parameter.



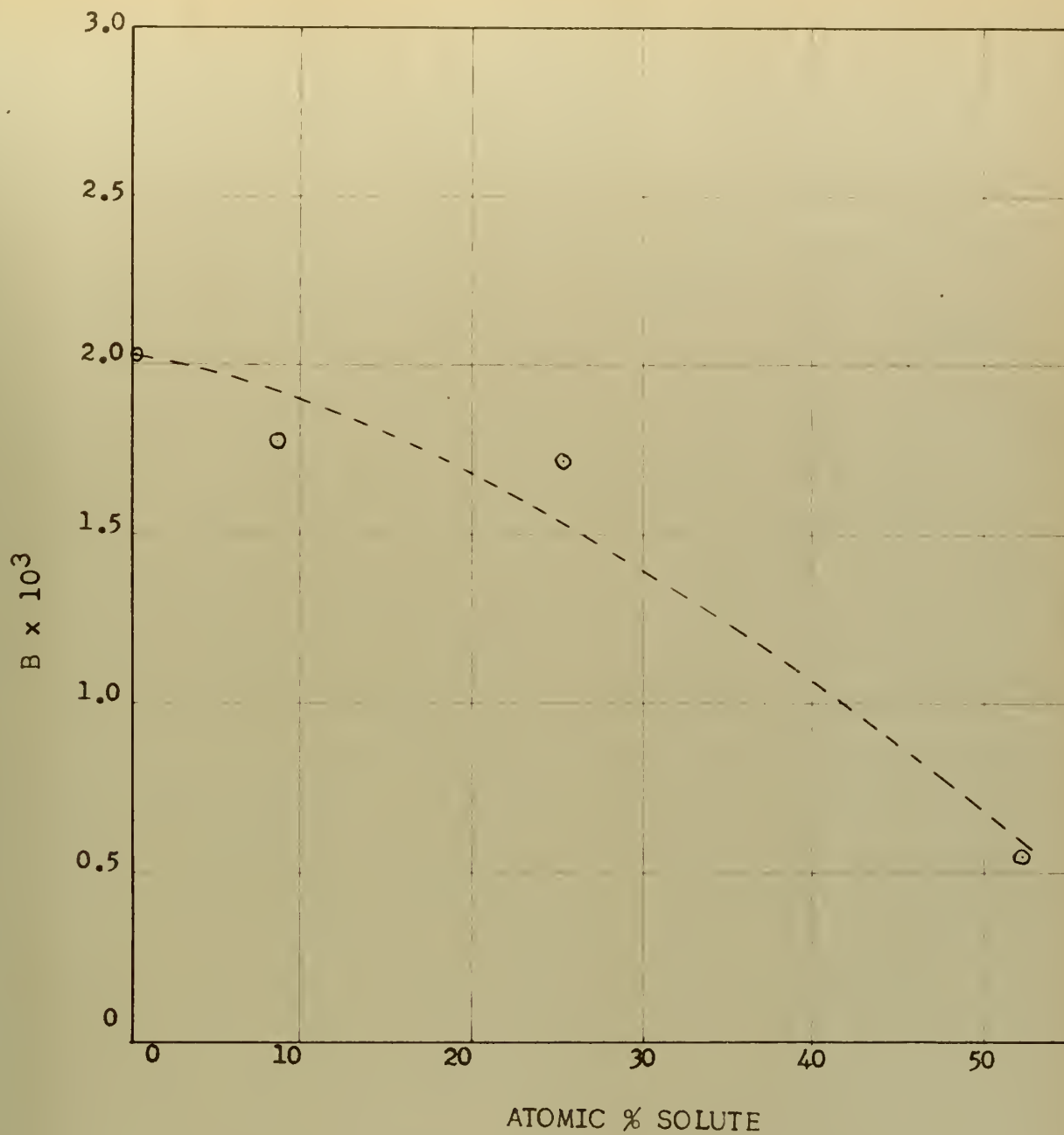


Fig. 10 EFFECT OF ALLOYING Mg IN Cd ON THE
B PARAMETER

CHAPTER VI

CONCLUSIONS

The analysis for this investigation depended on the validity of the expression: $\dot{\epsilon} = S e^{\frac{-\Delta H}{RT}} e^{\beta \sigma}$

This theory assumed obtaining the same value of $S e^{\frac{-\Delta H}{RT}}$ for each specimen, of a given alloy. The structure parameter S was maintained constant by the previously described procedure. Extensive investigations⁽¹³⁾ on a number of (alpha phase) solid solution alloys of aluminum showed that ΔH was insensitive to temperature, stress and structure. In addition, analyses by Sherby and Dorn⁽¹³⁾ of creep data found in the literature illustrated that ΔH was insensitive to at least temperature and stress for a wide variety of metals. Thus, it was felt justifiable to assume that a constant value of ΔH existed for each composition. The term $(S e^{\frac{-\Delta H}{RT}})$ was, therefore, considered a constant K . Thus, it was assumed that the creep rate is related to the stress in accordance with $\dot{\epsilon} = K e^{\beta \sigma}$

The results of this investigation show, that for a series of cadmium-magnesium alloys, that the B parameter of this relation is independent of the precreep strain and therefore of the structure developed during creep. It was found that the B parameter for cadmium decreased with increased alloying content of magnesium. For the cadmium-magnesium alloys tested, it is concluded that the B parameter is a function of the material



only.

A comparison between the effect of several elements in aluminum on the tensile strength of aluminum and the B parameter showed that those elements which increase the low temperature tensile strength also had the greatest effect on the B parameter. The order of magnitude of the changes in electronegativity* and atomic radii between cadmium and magnesium is of the same order of magnitude as between aluminum and those solute elements which show the greatest effect. Yet, the addition of magnesium to cadmium is only approximately 1/100 as effective in decreasing B, compared to the effect in the aluminum solid solutions. The aluminum alloys all had limited solubility ranges. The cadmium-magnesium binary system has an unlimited solid solution range. The fact that this system does not show any tendency toward compound formation (except ordering at lower temperatures) strongly indicates that the interaction between cadmium and magnesium is much less than suggested by the above factors of electronegativity, etc. It would be of interest to test these series in tension to further check on the difference in the effects of magnesium on cadmium as compared to the previously investigated alloys (14).

*As exhibited by the electrochemical potential in aqueous solutions



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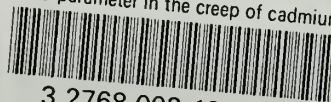
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